

TITLE OF THE INVENTION

LONG-LIFE HEAT-RESISTING LOW ALLOY STEEL WELDED
COMPONENT AND METHOD OF MANUFACTURING THE SAME

CROSS-REFERENCE TO RELATED APPLICATIONS

5 This application is based upon and claims the
benefit of priority from the prior Japanese Patent
Applications No. 2002-322969, filed November 6, 2002;
and No. 2003-357568, filed October 17, 2003, the entire
contents of both of which are incorporated herein by
10 reference.

BACKGROUND OF THE INVENTION

1. Field of the Invention

15 The present invention relates to a heat-resisting
low alloy steel welded component of a structure used
under a high-temperature and high-pressure atmosphere
such as a thermal power plant or petrochemical plant.
The invention also relates to a method of manufacturing
a welded component of heat-resisting steel used in
facilities that use in many occasions a thick structure
20 of a ferritic heat-resisting low alloy steel.

2. Description of the Related Art

25 Recently, boilers are employed under a higher
temperature and pressure, and their capacities are
increasing. With this development, thermal power
plants that can operate under a steam pressure of
316 kg/cm² and a steam temperature of 566°C and have
a large capacity that is equivalent to a 700,000 to

1,000,000 kW-class plant are constructed at present. The steel pipe material used for such a boiler is mainly divided into carbon steel or low-alloy steel, and high alloy steel, a typical example of which is austenitic stainless steel. These materials are carefully selected in accordance with the temperature and pressure of the section where the material is used. For example, for a hot reheat steam pipe or main steam pipe in which the temperature becomes 450°C or higher when in use, Mo steel or Cr-Mo steel, which has an excellent anti-oxidation property and heat resisting property, is used.

At present, for a large-diameter pipe such as a main steam pipe or hot reheat steam pipe that connects a boiler and a turbine, a thick pipe made of Cr-Mo low alloy steel is employed since these pipes have to stand a high internal pressure and temperature. In particular, in the thermal power plants in which a high-temperature and high-pressure steam is generated, such ferritic low alloy heat-resisting steel that is defined by JIS G 3458 (1988) is used for the pipes.

In welding of a steam pipe, a weld with a high joint efficiency is used. When such a steam pipe is used continuously under a high temperature and high pressure for a long period of time, various creep damages are created at the weld zone. There are four types of the creep damages as shown in FIG. 1. More

specifically, there are a damage of type 1 created in
a weld metal (deposit) 3, a damage of type 2 created in
a section from a weld metal 3 to a heat-affected zone
(HAZ) 2, a damage of type 3 created from a weld metal
5 side of the HAZ 2, and a damage of type 4 created in
a base metal 1 side of the HAZ 2. In particular, of
these, the type 4 damage, which is created in a fine
grained region of the HAZ, causes to significantly
shorten the lifetime of the weld zone of the pipe, and
10 therefore this type of damage is perceived as a problem
in many countries. The type 4 damage is discussed in,
for example, a non-patent document, "Review of Type IV
Cracking" written by F.V. Ellis and R. Viswanathan
(ASME PVP vol. 380, July 1998).

15 At present, a heat-resisting steel pipe is
manufactured in a procedure shown in FIG. 8B. That is,
a steel lumber is carried out an acceptance test.
The tested steel lumber is marked and then cut into
steel pieces (S11). Then, steel pieces are bent by
20 hot working (S12). Another marking is carried out on
the pieces and then cut (S13). The cut pieces are
subjected to a heat treatment (S14). Subsequently,
the treated pieces are subjected to a mechanical edge
preparation (15). The prepared pieces are aligned
25 edges each other (S16) and then welded together by
a submerged arc welding method (S17).

When manufacturing a heat-resisting steel pipe by

processing ferritic low alloy steel with the above-described procedure, it is difficult to avoid the occurrence of a creep damage in the HAZ. Especially, in the manufacture of a thick steel pipe having
5 a thickness of 25 to 150 mm, the submerged arc welding is carried out for 5 to 50 layers. Therefore, the base metal situated near a weld zone is exposed to multiple heat cycles, in which quick heating and quick cooling are repeatedly carried out. For this reason, the
10 grains become finer and finer in the HAZ. Further, in the case where, for example, such heat-resisting steel pipes are used to build a large-diameter pipe for a thermal power plant, these heat-resisting steel pipes are exposed to a high-temperature and high-pressure
15 environment such as a temperature of 538 to 566°C and a pressure of 169 to 316 kg/cm² due to the steam flowing through the large-diameter pipe. Further, a stress of 4 kg/mm² load on the pipe in its circumferential direction. Therefore, in the HAZ fine grained region,
20 a creep damage of the type 4 easily occurs, thereby shortening the lifetime of the component. The damaged section must be repaired, or if the damage is too severe to repair, the pipe itself must be replaced, costing a great amount of expense.

BRIEF SUMMARY OF THE INVENTION

The present invention has been proposed as a solution to the above-described drawback of the

conventional technique, and the object thereof is to provide a long-life heat-resisting low alloy steel welded component that does not easily cause a creep damage in a heat-affected zone and a method of manufacturing the same.

According to an aspect of the present invention, there is provided a method of manufacturing a long-life heat-resisting low alloy steel welded component including the steps of subjecting a base metal containing, at % by weight, C: 0.15% or less, Si: 0.5% or less, Mn: 0.3 to 0.8%, Cr: 1.9 to 2.6%, Mo: 0.87 to 1.20%, and a balance of iron and unavoidable impurities, to a hot working (S3), to a heat treatment (S5), and then to a welding (S8), the method being characterized by normalizing the base metal once or more times before the welding (S8) in addition to the hot working (S3).

The method of the present invention can be applied to a base metal containing, at % by weight, Mn: 0.3 to 0.6% and Mo: 0.87 to 1.13%. In this case, it is preferable that the normalizing of the base metal should be carried out at least twice.

According to another aspect of the present invention, there is provided a method of manufacturing a long-life heat-resisting low alloy steel welded component including the steps of subjecting a base metal containing, at % by weight, C: 0.04% to 0.10%,

Si: 0.5% or less, Mn: 0.1 to 0.6%, Cr: 1.9 to 2.6%,
Mo: 0.05 to 0.3%, V: 0.20 to 0.30%, Nb: 0.02 to 0.08%,
W: 1.45 to 1.75%, B: 0.0005 to 0.006% and a balance of
iron and unavoidable impurities, to a hot working, to
5 a heat treatment, and then to a welding, the method
being characterized by normalizing the base metal once
or more times before the welding in addition to the hot
working.

According to still another aspect of the present
10 invention, there is provided a method of manufacturing
a long-life heat-resisting low alloy steel welded
component including the steps of subjecting a base
metal containing, at % by weight, C: 0.2% or less,
Si: 1.0% or less, Mn: 0.3 to 0.9%, Cr: 0.3 to 1.5%,
15 Mo: 0.4 to 0.7%, and a balance of iron and unavoidable
impurities, to a hot working, to a heat treatment, and
then to a welding, the method being characterized by
normalizing the base metal once or more times before
the welding in addition to the hot working.

20 The method of the present invention can be applied
to a base metal containing, at % by weight, Mn: 0.3 to
0.6%, Cr: 0.5 to 1.5% and Mo: 0.40 to 0.65%.

The method of the present invention can be applied
to the base metal further containing, at % by weight,
25 V: 0.22 to 0.50%.

The method of the present invention can be applied
to a base metal that has been subjected to annealing or

normalizing and tempering.

It is preferable that the method of the present invention should be carried out by subjecting the base metal to the normalizing and then to the hot working in
5 a normalizing temperature range.

The long-life heat-resisting low alloy steel welded component according to the present invention can be manufactured by the above-described methods.

The welded component of the present invention can
10 be applied to at least one of longitudinal joint and circumferential joint of pipes, vessel, valve casing and branch pipes that are used under a high-temperature and high-pressure steam atmosphere at a temperature of 450°C or higher.

15 Further, the form of the base metal may be a plate, pipe, forging and casting.

The inventors of the present invention carried out intensive studies on creep damages of type 4, and discover the following facts. The present invention
20 has been achieved based on these findings.

The mechanism of occurrence of creep damages of type 4 is schematically illustrated in FIG. 2. Further, the results of observation of samples extracted from a site where damages of type 4 were
25 created under a transmission electron microscope (TEM) or scanning electron microscope (SEM) are shown in the photographs of FIGS. 3 to 7.

(I) At an former austenite (γ) grain boundary 11 of a to-be-welded material, coarse carbide 12 resides as shown at (a) in FIG. 2. Note that there are ferrite grains (not shown at (a) in FIG. 2) within the former austenite grains. As the material is exposed to heat cycles during welding, the reduction of size of bainite grains 13, 14 is promoted and at the same time, coarse carbide 12 is decomposed and solid-dissolved into specific bainite 14. Thus, bainite fine grains 14 having an extremely dense precipitation of carbide are arranged along the former austenite grain boundary 11 as shown at (b) in FIG. 2. Particularly, in the case of jointing thick steel pipes, the submerged arc welding with a large amount of heat input, is employed and the pipes are exposed to multiple heat cycles repeated in 5 to 50 times. Therefore, the bainite texture 13, 14 located close to the weld zone is decomposed into finer grains due to quick heating and quick cooling and thus high-carbide bainite fine grains 14 are formed along the grain boundary 11.

(II) Next, an enlarged schematic view of a bainite fine grain 14 shown at (b) in FIG. 2 is illustrated at (c) in FIG. 2. Here, due to the heat treatment carried out after the welding and also the actual use under a high temperature and high pressure, coarse carbides 16 are re-precipitated along a grain boundary 15 as well as within a grain 14a of bainite fine grain 14.

As a result, a plate-like agglomeration region is created as shown at (d) in FIG. 2. FIGS. 3 to 5 are TEM photographs corresponding to the illustration shown at (d) in FIG. 2. FIG. 3 is a TEM photograph taken
5 under a magnification of about 500 times, FIG. 4 is a photograph under a magnification of about 3,000 times, and FIG. 5 is a photograph taken under a magnification of about 10,000 times. In particular, a plate-like agglomeration region of the coarse carbide 16 and
10 unagglomerated spherical carbides 16s that are present around the plate-like agglomeration region can be observed in FIGS. 4 and 5.

(III) Further, when the welded component is used under conditions of a high temperature and high
15 pressure for a long period of time, creep damages significantly develop in an interface between a coarse carbide 16 re-precipitated along the grain boundary 15 of the bainite fine grain 14 and a matrix 14a. As a result, creep voids 17 are generated particularly in
20 that area, as shown at (e) in FIG. 2. As these creep voids 17 grow and become connected together, a grain boundary separation 18 is created as shown at (f) and (g) in FIG. 2. FIG. 6 is an SEM photograph corresponding to the illustration shown at (e), (f) and (g) in
25 FIG. 2. It is observed from FIG. 6 how dark portions (creep voids 17) are created in the grain boundary and these creep voids 17 are connected together into

a grain boundary separation 18.

(IV) Further, the separated portion 18 at the grain boundary grows as it is connected each other, and these portions develop into a crack 19 created along the former austenite (γ) grain boundary 11 as shown at (h) in FIG. 2. FIG. 7 is an SEM photograph corresponding to the illustration shown at (h) in FIG. 2. It is observed from FIG. 7 how the crack 19 is created in the welded component.

Due to the mechanism described in the above (I) to (IV), type 4 crack occurs on the base metal side of the HAZ.

Based on the findings described above, the present invention aims to reduce the amount of coarse carbides, which cause to damages of type 4, remaining in former austenite (γ) grain boundaries of a welded component (base metal). With the invention, the creep lifetime of the weld zone of the heat-resisting steel can be significantly improved. More specifically, before the base metal is welded, it is subjected to normalizing once or more times, thereby reducing the amount of coarse carbides.

Additional objects and advantages of the invention will be set forth in the description which follows, and in part will be obvious from the description, or may be learned by practice of the invention. The objects and advantages of the invention may be realized and

obtained by means of the instrumentalities and combinations particularly pointed out hereinafter.

BRIEF DESCRIPTION OF THE SEVERAL VIEWS OF THE DRAWING

The accompanying drawings, which are incorporated
5 in and constitute a part of the specification,
illustrate presently preferred embodiments of the
invention, and together with the general description
given above and the detailed description of the
preferred embodiments given below, serve to explain the
10 principles of the invention.

FIG. 1 is a schematic diagram illustrating types
of creep damage in a weld zone;

FIG. 2; (a) in FIG. 2 is a schematic diagram
illustrating residual coarse carbides, (b) and (c) in
15 FIG. 2 are schematic diagrams illustrating fine bainite
grains which have dense precipitation of carbides, (d)
in FIG. 2 is a schematic diagram illustrating a carbide
agglomeration region, (e) in FIG. 2 is a schematic
diagram illustrating formation of creep voids, (f) and
20 (g) in FIG. 2 are schematic diagrams illustrating grain
boundary separation, and (h) in FIG. 2 is a schematic
diagram illustrating occurrence of a crack;

FIG. 3 is a TEM photograph showing a carbide
agglomeration region of a welded component, taken under
25 a magnification of about 500 times;

FIG. 4 is a TEM photograph showing the carbide
agglomeration region at the same site as that shown

in FIG. 3, taken under a magnification of about 3,000 times;

FIG. 5 is a TEM photograph showing the carbide agglomeration region at the same site as that shown in FIG. 3, taken under a magnification of about 10,000 times;

FIG. 6 is an SEM photograph showing creep voids and separated portions along a bainite grain boundary of a welded component, taken under a magnification of about 2,500 times;

FIG. 7 is an SEM photograph showing a portion of the welded component where a crack is created, taken under a magnification of about 500 times;

FIG. 8A is a flowchart illustrating a method of manufacturing a welded component according to an embodiment of the present invention, and FIG. 8B is a flowchart illustrating a conventional method of manufacturing a welded component; and

FIG. 9 is a TEM photograph showing an observation sample of a welded component according to Reference Example, under a magnification of about 500 times.

DETAILED DESCRIPTION OF THE INVENTION

An embodiment of the manufacturing method according to the present invention will now be described in detail with reference to FIG. 8; however the present invention is not limited to this embodiment.

As shown in FIG. 8A, a steel lumber 21 is carried out an acceptance test (S1). The steel lumber 21 is, as defined, for example, by JIS G 3458, either a material that has been subjected to normalizing and tempering once (NT material) or a material that has been subjected to annealing once (annealed material). The steel lumber 21 is marked and then cut into steel pieces 22. The steel pieces 22 are subjected normalizing at a predetermined temperature (S2). Subsequently, they are bent by hot working with use of a recess type mold and a projection type mold at a temperature within a range of 900 to 1,000°C into bent steel pieces 22a (S3). In the case where the hot press bending is not carried out to follow the normalizing, the steel pieces are heated up to a temperature within a range of 900 to 1,000°C, and then subjected to hot press bending.

Next, again, the bent steel pieces 22a are marked and then cut (S4). The cut pieces are subjected to a heat treatment within a predetermined temperature range in a furnace 25 (S5). After the heat treatment, two or more bent steel pieces are placed on an edge rest 26 and then subjected to a mechanical edge preparation with a cutting edge 27. After that, the bent steel pieces 22a are aligned their edges each other (S7) and they are welded together by hands from the first to second or third layers. Then, they are

subjected to a full-automatic submerged arc welding by 5 to 50 passes, and thus a linear joint steel pipe having a longitudinal joint is prepared (S8).

Conventionally, when welding heat-resisting steel,
5 immediately after the steel material 21 is carried out an acceptance test (S11), it is heated up to a temperature within a range of 900 to 950°C, and then subjected to hot press bending (S12). However, with this operation, coarse carbides reside in a former
10 austenite grain boundary, and thus damages of type 4 are created. In order to avoid this, the present invention provides a step of normalizing (S2) after the carrying out an acceptance test (S1), and then the hot working (S3) is carried out. With this operation, the
15 coarse carbides remaining in the former austenite grain boundary are solid-dissolved and diffused into the matrix. In this manner, the generation of the creep voids can be prevented and thus the type 4 damages can be avoided.

20 The following are explanations of the reasons for providing various limitations in the present invention. It should be noted the expression "high temperature" used in this specification indicates mainly a temperature in a range of 450 to 650°C unless the
25 temperature is particularly specified.

(Compositions)

Note that unless particularly specified,

the expression "%" indicates % by weight.

(1) C: 0.2% or less

C contributes to an increase in the strength of the steel. However, if C is added to exceed 0.2%,
5 a harmful effect of S is promoted, and the grain boundary separation is easily induced. At the same time, it is possible that the toughness or weldability is deteriorated. For this reason, the C content should be defined to 0.2% or less, preferably 0.15% or less,
10 or more preferably 0.10% or less. Meanwhile, at least 0.04% of C is required in the steel so as to assure a high-temperature strength, and therefore it is preferable that the C content should be 0.04% or more.

(2) Si: 1.0% or less

15 Si contributes to an increase in the strength of the low-alloy steel and it has a deoxidation effect. However, if Si is added to exceed 1.0%, the HAZ toughness and weldability are deteriorated. For this reason, the Si content should be defined to 1.0% or
20 less, preferably 0.5% or less. Meanwhile, if the Si amount is excessively small, it is possible that the amount of oxide-based inclusions in the steel is increased, thereby deteriorating the strength. Therefore, it is preferable that the Si content should
25 be 0.15% or more.

(3) Mn: 0.3 to 0.9% or 0.1 to 0.6%

Mn is an element effective for preventing the

grain boundary separation, serves to improve the strength and toughness of the steel lumber and further serves to suppress the generation of FeS. MnS, which is created when Mn and S bond together, has a high melting point, and therefore Mn functions to prevent the weakening of the grain boundary, which is caused by the presence of S. For this reason, the Mn content is defined to be 0.3% or more. Meanwhile, if Mn is added to exceed 0.9%, the hardenability is increased. Thus, a hardened layer is formed during the welding and therefore the weld cracking parameter P_{CM} is increased. For this reason, the Mn content is defined to 0.3 to 0.9%, preferably 0.3 to 0.8% and more preferably to 0.3 to 0.6%.

Alternatively, when W and B, which are effective for improving the strength of the steel lumber, are both added, the Mn content may be reduced to a range of 0.1 to 0.6%.

(4) Cr: 0.3 to 1.5% or 1.9 to 2.6%

Cr is an element effective for improving the oxidation resistance, but when its effect is low when the Cr content is less than 0.3%. However, if Cr is added to exceed 2.6%, the weldability and HAZ toughness are deteriorated.

Particularly, in the case where the Mo content is set to 0.87 to 1.20%, it is possible to significantly suppress the graphitizing of the pipe, which is exposed

to a high temperature for a long time while in use, if the Cr content is set to 1.9% or more. In the case where the Mo content is set to 0.4 to 0.7%, the above-described effect can be sufficiently obtained if the Cr content is set to 0.3 to 1.5%, more preferably to 0.5 to 1.5%.

(5) Mo: 0.05 to 0.3%, 0.4 to 0.7% or 0.87 to 1.20%

Mo is an element effective for improving the hardenability and increasing the resistance to temper softening, and thus it is effective for increasing the high temperature strength. This element is most effective for increasing the creep strength. However, when the Mo content is less than 0.4%, its advantageous effects cannot be sufficiently exhibited. On the other, if Mo is added to exceed 1.20%, the weldability is deteriorated. Further, carbides are deposited and therefore the yield ratio is increased.

Especially, in the case where the Cr content is set within 1.9 to 2.6%, the creep property can be further improved by setting the Mo content to 0.87% or more. Thus, the material can be used at a higher temperature. Therefore, as described before, when the Cr content is set to 0.3 to 1.5%, the Mo content should be set to 0.4 to 0.7%, more preferably to 0.40 to 0.65%. Alternatively, when the Cr content is set to 1.9 to 2.6%, the Mo content should be set to 0.87 to 1.20%, more preferably to 0.87 to 1.13%.

Alternatively, when W and B are added each in an appropriate amount, a similar effect to that of the addition of Mo can be obtained. When W and B are both added, the Mn content may be reduced to a range of 0.05 to 0.3%.

(6) V: 0.22 to 0.50% (0.20 to 0.30%)

V, when a small amount is added, serves to improve the hardenability and increase the resistance to temper softening. Further, when V is added to a Cr-Mo steel to have a content of 0.22%, the high temperature strength is increased and the creep strength is improved. However, when the V content is less than 0.22%, its advantageous effects cannot be sufficiently exhibited. On the other, if V is added to exceed 0.50%, the weldability is deteriorated. For this reason, in the case where V is added, the V content should preferably be set to 0.22 to 0.50%. Especially, in the case of manufacturing a thick steel pipe, which requires even a higher weldability, it is preferable that the V content should be set to 0.22 to 0.32%.

Note that in the case of a W, B and Nb-added steel, the V content is set to 0.20 to 0.30%.

(7) Nb: 0.02 to 0.08%

Nb has an effect of reducing the size of grains and making the material to have a higher temperature point at which grains are made coarse. Therefore, with an appropriate amount of Nb added, the strength

can be increased, and the ductility and toughness can be improved. When the Nb content is less than 0.02%, the above-described effect cannot be sufficiently exhibited. On the other hand, when Nb is added to exceed 0.08%, δ -ferrite is generated and further coarse carbides are created, thereby deteriorating the ductility and toughness. Further, the agglomeration of carbides takes place quickly, possibly causing a decrease in the creep rupture strength. For this reason, in the case where Nb is added, its content is set to 0.02 to 0.08%.

(8) W: 1.45 to 1.75%

W can serve as a substituting element for Mo, and when an appropriate amount of this element is added, it serves to improve the high-temperature creep strength. However, when the W content is less than 1.45%, its advantageous effects cannot be sufficiently exhibited. Here, W is an expensive metal, and therefore if it is added the content should be 1.75% or less for the economical reason. Thus, in the case where the Mo content is 0.05 to 0.3%, it is preferable that W should be added to have a content of 1.45 to 1.75%.

(9) B: 0.0005 to 0.006%

B serves to strengthen the grain boundaries and increase the high-temperature strength, when an appropriate amount of this element is added. However, when the B content is less than 0.0005%, its

advantageous effects cannot be sufficiently exhibited. On the other hand, when B is added excessively to have a content of more than 0.006%, not only the creep rupture strength is decreased but also the creep rupture ductility is lowered, thereby possibly causing notch brittleness. Therefore, in the case where the W content is 1.45 to 1.75%, it is preferable that B should be added to have a content of 0.0005 to 0.006%, more preferably 0.0005 to 0.003%.

10 (10) Ni: 0.30% or less

Ni serves to improve the quench hardenability. With a small amount of Ni added, an excellent advantageous effect can be obtained especially in the case of manufacturing a thick steel pipe. However, if Ni is added to exceed 0.30%, not only an excessive hardenability is obtained but also it is possible to deteriorate the stability of ferrite. Therefore, if Ni is contained, its content should preferably be set to 0.30% or less.

20 (11) Cu: 0.30% or less

Cu has an effect of providing an corrosion resistance. However, if Cu is added to exceed 0.30%, it may cause a red shortness or Cu-crack by welding. Therefore, if Cu is contained, its content should preferably be set to 0.30% or less.

(12) Sn: 0.030% or less

Sn contributes, if added in a small amount,

to improvement of the fatigue limit and corrosion resistance. However, if Sn is added to exceed 0.030%, it may cause deterioration in the creep property, the hot-workability, the toughness or the like. Therefore, the Sn content should preferably be reduced as much as possible, and more preferably, it should be set to 0.030% or less.

(13) sol.Al: 0.030% or less

Acid-soluble aluminum (sol. Al) serves as an effective deoxidizer. However, if sol. Al is added to exceed 0.030%, the toughness is decreased. Therefore, if sol. Al is contained, its content should preferably be set to 0.030% or less, or more preferably to 0.025% or less. Further, when the sol. Al content is set to 0.020% or less, the toughness can be further improved.

(14) Unavoidable Impurities

P: 0.035% or less

P is an impurity contained in steel, and it intensifies the segregation and remarkably promotes the temper embrittlement. For this reason, the P content should preferably be set to 0.035% or less, more specifically to 0.030% or less.

S: 0.035% or less

S is an impurity contained in steel. FeS is weak and crystallized in grain boundaries, thereby embrittling the steel lumber. For this reason, the S content should preferably be set to 0.035% or less,

more preferably to 0.030% or less, and even more preferably to 0.010% or less.

Next, standard steel types that can be used as a base metal of the present invention will now be listed.

5 First, the steel types for plates will be listed.
(These are cited from Reference Document 1, "Handbook of Comparative World Steel Standards Second Edition" edited by John E. Bringas, 2002; ASTM Stock Number: DS67A, ASTM INTERNATIONAL, Chapter 4 Pressure Vessel
10 Steel Plates.)

ASTM A 387/A 387M Gr.2, Cl.1 K12143; JIS G 4109
SCMV 1 Div.1 and SCMV 1 Div.2; ASTM A 387/A 387M Gr.2,
Cl.2 K12143

15 ASTM A 387/A 387M Gr.12, Cl.1 K11757; JIS G 4109
SCMV 2 Div 1; ASTM A 387/A 387M Gr.12, Cl.2 K11757;
JIS G 4109 SCMV 2 Div 2; EN 10028-2 13CrMo4-5 1.7335;
ISO 9328-2 14CrMo4 5

JIS G 4109 SCMV 3 Div 1; ASTM A 387/A 387M Gr.11,
Cl.1 K11789 and Gr.11, Cl.2 K11789; JIS G 4109 SCMV 3
20 Div 2

JIS G 4109 SCMV 4 Div 1; ASTM A 387/A 387M 22,
Cl.1 and 22L, Cl.1 K21590; EN 10028-2 10CrMo9-10
1.7380; ISO 9328-2 13CrMo9 10 T1; ASTM A 387/A 387M 22,
Cl.2 K21590; JIS G 4109 SCMV 4 Div 2; EN 10028-2
25 11CrMo9-10 1.7383; ISO 9328-2 13CrMo9 10 T2

JIS G 4109 SCMV 5 Div 1; ASTM A 387/A 387M Gr.21,
Cl.1, Gr.21L, Cl.1 and Gr.21, Cl.2 K31545; JIS G 4109

SCMV 5 Div 2

JIS G 4109 SCMV 6 Div 1; ASTM A 387/A 387M Gr.5, Cl.1 and Gr.5, Cl.2 K41545 S50100 and S50200; JIS G 4109 SCMV 6 Div 2

5 Further, for a steel plate, KA-SCMV4J1 specified by Code for Thermal Power Generation Facilities (to be abbreviated as "CTPGF" hereinafter) of JSME standard can be also used. (Reference document 2: "Code for Thermal Power Generation Facilities" 1999 edition and
10 2002 additional edition edited by Japan Electrical Engineering Specification Committee, published by Japan Society of Mechanical Engineers, JESC 10005 (2002) extracted from <TABLE II-1-1 Notes> 5)

Further, for steel plates, JIS G 4110 SCMQ4E,
15 SCMQ4V and SCMQ5V can be employed as well.

Examples of the steel pipe will now be listed.
(These are extracted from Reference Document 1, Chapter 5 Steel Tubes and Pipes.)

AFNOR NF A 49-215 TU 15 D 3; AFNOR NF A 49-245 TS
20 15 D 3

ASTM A 213/A 213M T2 K11547; ASTM A 250/A 250M T2 K11547; JIS G 3462 STBA20; AFNOR NF A 49-215 TU 15 CD 2-05; AFNOR NF A 49-245 TS 15 CD 2 05; ISO 2604-II TS 33

25 ASTM A 213/A 213M T12 K11562; ASTM A 250/A 250M T12 K11562; JIS G 3462 STBA22; JIS G 3467 STFA22; AFNOR NF A 49-245 TS 15 CD 4 05; BSI BS 3059-2 620-460;

BSI BS 3606 620; DIN 28180 13CrMo44 1.7335; ISO 2604-II
TS 32; ISO 2604-III TW 32

ASTM A 213/A 213M T11 K11597; ASTM A 250/A 250M
T11 K11597; JIS G 3462 STBA23; JIS G 3467 STFA23;

5 BSI BS 3606 621; AFNOR NF A 49-215 TU 10 CD 5-05

ASTM A 213/A 213M T22 K21590; ASTM A 213/A 213M
T23; ASTM A 250/A 250M T22 K21590; JIS G 3462 STBA24;
JIS G 3467 STFA24; BSI BS 3059-2 622-490; BSI BS 3606
622; AFNOR NF A 49-215 TU 10 CD 9-10; ISO 2604-II TS 34

10 ASTM A 213/A 213M T5 K41545 and T5b K51545; JIS
G 3462 STBA25; JIS G 3467 STFA25; BSI BS 3606 625;
AFNOR NF A 49-215 TU Z 10 CD 5 05; ISO 2604-II TS 37

AFNOR NF A 49-213 TU 15 D 3; AFNOR NF A 49-220 TU
15 D 3; AFNOR NF A 49-253 TS 15 D 3

15 ASTM A 335/A 335M P2 K11547; JIS G 3458 STPA20;
AFNOR NF A 49-213 TU 15 CD 2-05; AFNOR NF A 49-220 TU
15 CD 2-05; AFNOR NF A 49-243 TS 15 CD 2-05; AFNOR NF A
49-253 TS 15 CD 2-05

20 BSI BS 3604-1 AMD 2 660(0.5Cr-0.5Mo-0.25V); DIN
17175 14MoV63 1.7715

ASTM A 335/A 335M P12 K11562; JIS G 3458 STPA22;
BSI BS 3604-1 AMD 2 620-440(1Cr-0.5Mo); BSI BS 3604-2
620; DIN 17175 13CrMo44 1.7335; AFNOR NF A 49-213 TU 13
CD 4-04; AFNOR NF A 49-219 TU 13 CD 4-04 and TU 10 CD
25 5-05; AFNOR NF A 49-220 TU 10 CD 5-05; AFNOR NF A
49-243 TS 15 CD 4-05

ASTM A 335/A 335M P11 K11597; JIS G 3458 STPA23;

BSI BS 3604-1 AMD 2 621(1.25Cr-0.5Mo); BSI BS 3604-2
621; AFNOR NF A 49-213 TU 10 CD 5-05; AFNOR NF A 49-220
TU 13 CD 4-04(1)

ASTM A 335/A 335M P22 K21590; JIS G 3458 STPA24;
5 BSI BS 3604-1 AMD 2 622(2.25Cr-1Mo); BSI BS 3604-2 622;
DIN 17175 10CrMo9 10 1.7380; AFNOR NF A 49-213 TU 10 CD
9-10; AFNOR NF A 49-219 TU 10 CD 9-10; AFNOR NF A
49-220 TU 10 CD 9-10; AFNOR NF A 49-253 TS 10 CD 9-10

ASTM A 335/A 335M P5 K41545; JIS G 3458 STPA25;
10 BSI BS 3604-1 AMD 2 625(5Cr-0.5Mo); AFNOR NF A 49-213
TU Z 12 CD 05-05; AFNOR NF A 49-219 TU Z 12 CD 05-05;
AFNOR NF A 49-253 TS Z 10 CD 5-05

With regard to the steel pipes, those listed below
can be employed as well. (Extracted from Reference
15 Document 2 <TABLE II-1-1 Notes> 8 and 14.)

CTPGF KA-STBA21 of JSME standard; CTPGF KA-STBA24
E-G of JSME standard; CTPGF KA-STBA24J1 of JSME
standard

CTPGF KA-STPA21 of JSME standard; CTPGF
20 KA-STPA24J1 of JSME standard

Further, the following steel pipe types are
also employable, that is, JIS G 5202 SCPH 21-CF and
SCPH 32-CF.

The steel types for forgings will now be listed.
25 (Reference Document 1, extracted from Chapter 6 Steel
Forgings.)

ISO 9327-2 16Mo3

JIS G 3203 SFVA F 2; ASTM A 182/A 182M F 2 K12122
EN 10222-2 13CrMo4-5 1.7335; ISO 9327-2 14CrMo4-5;
JIS G 3203 SFVA F 12; ASTM A 182/A 182M F 12, Cl 2
K11564; ASTM A 336/A 336M F12 K11564

5 JIS G 3203 SFVA F 11 A; ASTM A 182/A 182M F 11,
Cl 2 K11572; ASTM A 336/A 336M F11, Cl2 and F11, Cl3
K11572; JIS G 3203 SFVA F 11 B

JIS G 3203 SFVA F 22 A; ASTM A 182/A 182M F 22,
Cl 1 K21590; ASTM A 336/A 336M F22, Cl 1 K21590;
10 EN 10222-2 11CrMo9-10 1.7383; ISO 9327-2 13CrMo9-10;
ASTM A 182/A 182M F 22, Cl 3 K21590; ASTM A 336/A 336M
F22, Cl3 K21590; JIS G 3203 SFVA F 22 B

JIS G 3206 SFVCM F22B; ASTM A 508/A 508M 22, Cl 3
K21590; ASTM A 541/A 541M 22, Cl 3 K21390

15 JIS G 3206 SFVCM F22V; ASTM A 336/A 336M F22V;
ASTM A 541/A 541M 22V; ASTM A 182/A 182M F 22V K31835

JIS G 3203 SFVA F 21 A; ASTM A 336/A 336M F21, Cl1
K31545; ASTM A 182/A 182M F 21 K31545; ASTM A 336/A
336M F21, Cl3 K31545; JIS G 3203 SFVA F 21 B

20 JIS G 3206 SFVCM F3V; ASTM A 182/A 182M F 3V
K31830; ASTM A 336/A 336M F3V; ASTM A 508/A 508M 3V
K31830; ASTM A 541/A 541M 3V K31830

EN 10222-2 X16CrMo5-1 1.7366; JIS G 3203 SFVA F 5
A; ASTM A 336/A 336M F5 K41545; ISO 9327-2 X12CrMo5-1

25 JIS G 3203 SFVA F 5 B; ASTM A 182/A 182M F 5
K41545

ASTM A 336/A 336M F5A K42544; JIS G 3203 SFVA

F 5 C

ASTM A 182/A 182M F 5a K42544; JIS G 3203 SFVA
F 5 D; EN 10222-2 X16CrMo5-1 1.7366

With regard to the forging steel, CTPGF
5 KA-SFVAF22AJ1 of JSME standard can be employed as well.
(Extracted from Reference Document 2 <TABLE II-1-1
Notes> 18.)

The steel types in castings will now be listed.
(Reference Document 1, extracted from Chapter7 Steel
10 Castings.)

JIS G 5151 SCPH 11; ISO 4991 C28H

JIS G 5151 SCPH 21; ASTM A 217/A 217M WC6 J12072;
ISO 4991 C32H; EN 10213-2 G17CrMo5-5 1.7357

JIS G 5151 SCPH 23; ASTM A 389/A 389M C24 J12092;
15 ISO 4991 C35BH; EN 10213-2 G17CrMoV5-10 1.7706

JIS G 5151 SCPH 32; ASTM A 217/A 217M WC9 J21890;
ISO 4991 C34AH; ASTM A 487/A 487M 8Cl.ABC J22091;
EN 10213-2 G17CrMo9-10 1.7379; ISO 4991 C34BH

JIS G 5151 SCPH 61; ASTM A 217/A 217M C5 J42045;
20 EN 10213-2 GX15CrMo5 1.7365; ISO 4991 C37H

Of these steel types, STPA 20, 22, 23 and 24
specified by JIS G 3458, CTPGF KA-STPA24J1 of JSME
standard, DIN17175 14MoV63 specified by German Standard
and Cr-Mo-V-based low-alloy steel specified by British
25 Standard BS3605-1 660 are picked up, and their chemical
compositions will be specified in TABLE 1 below.

TABLE 1

Steel type	C	Si	Mn	P	S	Cr	Mo	V	sol.Al	Others
JIS STPA20	0.10 to 0.20	0.10 to 0.50	0.30 to 0.60	0.035 or less	0.035 or less	0.50 to 0.80	0.40 to 0.65	-	-	-
JIS STPA22	0.15 or less	0.50 or less	0.30 to 0.60	0.035 or less	0.035 or less	0.80 to 1.25	0.45 to 0.65	-	-	-
JIS STPA23	0.15 or less	0.50 to 1.00	0.30 to 0.60	0.030 or less	0.030 or less	1.00 to 1.50	0.45 to 0.65	-	-	-
JIS STPA24	0.15 or less	0.50 or less	0.30 to 0.60	0.030 or less	0.030 or less	1.90 to 2.60	0.87 to 1.13	-	-	-
JSME CTPGF KA-STPA 24J1	0.04 to 0.10	0.50 or less	0.10 to 0.60	0.030 or less	0.010 or less	1.90 to 2.60	0.05 to 0.30	0.20 to 0.30	0.030 or less	Nb;0.02 to 0.08 N;0.030 or less W;1.45 to 1.75 B;0.0005 to 0.006
DIN17175 14MoV63	0.10 to 0.18	0.10 to 0.35	0.40 to 0.70	0.035 or less	0.035 or less	0.30 to 0.60	0.50 to 0.70	0.22 to 0.32	-	-
BS3604-1 660	0.10 to 1.05	0.10 to 0.35	0.40 to 0.70	0.030 or less	0.030 or less	0.30 to 0.60	0.50 to 0.70	0.22 to 0.28	0.02 or less	-

(Manufacturing Conditions)

A base metal having the above-specified composition is subjected to hot working and heat treatment, and then it is welded to prepare a component. Here, before the welding, the normalizing process is carried out once or more as a separate step from the hot working.

(a) Normalizing (S2)

Normalizing is carried out for a necessary number of times (once or more) so as to sufficiently reduce the amount of coarse carbides present in former austenite grain boundaries in steel, that is, to have the carbides solid-dissolved and diffused into the matrix. The number of times that the normalizing carried out should preferably be once or twice; however it may be varied widely depending on the thickness and the chemical compositions of the material. In the case of a particularly thick material or a material with a large amount of alloy elements (Cr, Mo, V) added (that is, for example, STPA24, DIN17175 14MoV63 and BS3604-1 660), the normalizing should preferably performed two or three or more times.

A ferritic heat-resisting low-alloy steel that can be used in the present invention is a normalized and tempered material (NT material), which is, when manufacturing its base metal (for example, a lumber 21 in FIG. 8A), heated up to a complete austenite region

and then tempered. In the case of the NT material, heating carried out during the manufacture of the base metal is regarded substantially as normalizing. Therefore, in the case of such a base metal, if the
5 normalizing is carried out at least once before the welding as a separate step from the hot working, it is practically assumed that a total of two or more times of normalizing are carried out. Thereby, the amount of coarse carbides can be reduced. In connection with an
10 annealed material, it is heated up to a quasi-austenite region (close to a temperature of complete austenite) when the base metal is manufactured. Therefore, if the normalizing is carried out at least once before the welding as a separate step from the hot working,
15 a sufficient effect can be obtained, but twice or more of normalizing is preferable.

The normalizing includes the following procedures. That is, a base metal is heated to an appropriate normalizing temperature of point Ac_3 or Ac_1 or higher
20 (for example, $Ac_3 + 50^{\circ}C$), and maintained at the temperature for a certain period of time. Then, the base metal is cooled down in the atmosphere. Alternatively, the base metal is cooled down to a temperature just above the Ar' point and maintained
25 at the temperature until the completion of the isothermal transformation, and then it is air-cooled. In the normalizing, if the maintained temperature is

excessively high, it is possible that the γ grains become coarse more than necessary, whereas if the maintained temperature is excessively low, coarse carbides may not be sufficiently solid-dissolved.

5 On the other hand, the maintained time period at the normalizing temperature is excessively long, it is possible that the γ grains become coarse more than necessary, thereby lowering the strength of the base metal. When the maintained time period is excessively
10 short, coarse carbides may not be sufficiently solid-dissolved. The maintained normalizing temperature can be set to 920 to 930°C in the case of, for example, STPA 24. The time period maintained at the normalizing temperature can be set to 0.5 to 1 hour in the case of,
15 for example, STPA 24. However, the appropriate maintaining temperature and maintaining time of the normalizing temperature vary depending on the type of steel employed. Therefore, it is preferable that the normalizing process should be carried out in accordance
20 with the specification of the steel type in each case.

It should be noted that when carrying out the normalizing, the temperature may be maintained isothermally or the heating carried out within the above temperature range may be maintained within the
25 above time range.

The normalizing of a base metal may be carried out at any timing if it is before the welding. However,

when the normalizing process is carried out after the hot working, some dimensional error may occur due to thermal strain, and therefore it is preferable that the normalizing should be carried out before the hot working.

The hot working, heat treatment and welding steps will now be described; however the present invention is not particularly limited to these, but the conventional techniques can be employed.

10 (b) Hot working (S3)

Examples of the hot working are press forming, rolling and bending. In these hot working techniques, the heating temperature, pressure and the like are not particularly limited, but it is preferable that the hot press bending should be carried out at a temperature within a range of 900 to 1,000°C continuously after the normalizing process. In the case where the hot working is not carried out continuously after the normalizing, it is preferable that the material should be heated to a temperature within a range of 900 to 1,000°C, and then the hot press bending should be carried out. It should be noted that if the hot working is carried out continuously after the normalizing process, the step of re-heating the material can be omitted, and therefore it is more efficient and thus more preferable.

The manufacturing method of the present invention can be applied to a welded steel pipe as well as a

seamless steel pipe. When the present invention is applied to a welded steel pipe, the hot working includes, for example, a step of hot pressing a hoop iron into a desired shape.

5 (c) Heat Treatment (S5)

In the case where, for example, STPA 20, 22, 23 or 24 defined by JIS is employed as the steel for a heat-resisting pipe, the material is subjected to normalizing and to hot working (press bending) first,
10 and then the material is marked and cut. After that, the material is tempered to a temperature within a range of 700 to 750°C.

On the other hand, in the case where, for example, a material called 14MoV63 by its steel type code
15 defined under DIN17175 is employed, the material is subjected to normalizing and to hot working (press bending) at a temperature within a range of 850 to 1100°C first, and then the material is marked and cut. After that, the material is annealed by air-cooling,
20 or by a liquid cooling method (such as oil quench hardening) if the material is thick, from a temperature within a range of 950 to 980°C. Subsequently, the material is tempered by maintaining the temperature within a range of 690 to 730°C for 30 minutes.

25 (d) Welding (S8)

The present invention can be applied not only to longitudinal joint or circumferential joint of

a longitudinal joint steel pipe as shown in FIG. 8, but also to circumferential joint of, for example, a spiral joint steel pipe or seamless steel pipe. Further, the present invention is applicable to longitudinal (seam) welding and circumferential welding carried out to prepare a steel pipe, but also to any of a pipe-to-end plate welding and pipe welding for boilers and heat exchanger.

The present invention can be applied to a heat-resisting low-alloy steel pipe used to make a large-diameter pipe employed in a thermal power plant or petrochemical plant. When such a large-diameter pipe is used in a situation where a high-temperature and high-pressure steam flows in the pipe, it may be exposed to an environment of a temperature of 450 to 600°C and a high pressure, in general cases. Especially, in the case of a thermal power plant, it may be exposed to an environment of a temperature of 538 to 566°C and a pressure of 169 to 316 kg/cm². Therefore, pipes having a thickness of 25 to 150 mm are used. For jointing thick pipes together, 5 to 50 layers, in particular, 30 to 40 layers must be welded by submerged arc welding in order to bury the edges. Thus, the base metal portion located close to the weld zone is exposed repeatedly to heat cycles of heating and cooling, and the width of the HAZ becomes about 2 to 3 mm. For this reason, creep damages of type 4,

in particular, are easily created in a thick pipe. For such thick pipes, the present invention can be effectively applied because it can reduce the amount of coarse carbides, which are origins of such creep damages, present along former austenite grain boundaries in steel, and thereby suppress the formation of a plate-like agglomeration region.

Next, examples of the present invention will now be described with reference to accompanying drawings; however, the present invention is not limited to these examples.

(Example 1)

As a sample steel, ASTM A387 Gr. D (, which is equivalent to ASTM A387/A 387M 22) corresponding to STPA 24 defined by JIS was prepared. The sample steel was subjected to hot press bending at a temperature of 900 to 930°C, and then marked and cut. The cut pieces were subjected to tempering for 4 hours at a temperature of 700 to 730°C. After that, the pieces were subjected to mechanical edge preparation, and they were aligned their edges each other. Then, the outer side and inner side of the bent site were welded by submerged arc welding. In this manner, a 90° -elbow plate bent longitudinal seam steel pipe having an outer diameter of 508 mm and a thickness of 106 mm was manufactured.

The obtained steel pipe was used as a main steam

pipe of a thermal power plant for about 145,000 hours to have it aged. The aging conditions are the maximum pressure of 177.5 kgf/cm² and the maximum temperature of 575°C and the average temperature of 566°C.

5 By using the base metal portion other than the weld zone of thus obtained after-use, i.e. damaged, pipe, a simulation test of the high temperature creep test was conducted in the following procedure. A summary of the simulation test will now be explained
10 with reference to FIG. 8A.

 The after-use, i.e. damaged, pipe was heated at a temperature of 1,000°C for 0.5 hours and then subjected to hot press to make the material into a plate shape. Then, the material was air-cooled and got a refreshed
15 material (, which is the first substantial normalizing) (S3). Two of such plate-like refreshed materials were prepared, and they were normalized at a temperature of 900°C for 2 hours and air-cooled (, which is the second substantial normalizing) (S2). After that, the
20 materials were tempered at a temperature of 750°C for 2 hours and air-cooled (S5). After that, a pair of these plate materials were mechanically beveled to form edges (S6). Then, they were aligned their edges each other (S7), and welded together by submerged
25 arc welding. In this manner, a sample weld having a thickness of 98 mm was manufactured (S8).

 The obtained sample weld was aged by applying

a fixed load of 400 kg/mm² onto a test piece with use of the method defined by JIS Z 2271 (1993). The test piece was obtained from a depth about two fifth of the thickness from the outer surface as it had the HAZ fine
5 grained region in the middle of the gauge. The aging conditions are a temperature of 680°C and the test was continued until the piece was ruptured. The rupture time of the test piece was about 220 hours. When converted under the actual use conditions, the rupture
10 time is about 240,000 hours.

After the aging, the test piece was observed under a transmission electron microscope (TEM) as to whether a type 4 damage has occurred or not. The result was as indicated in TABLE 2.

15 (Example 2)

A simulation test was carried out until the test piece was ruptured in a similar manner to that of Example 1 except that the aging temperature was set to 670°C. The rupture time of the test piece was about
20 380 hours. When converted under the actual use conditions, the rupture time is about 240,000 hours. Then, the test piece was observed as to whether a type 4 damage has occurred or not. The result was as indicated in TABLE 2.

25 (Example 3)

A simulation test was carried out until the test piece was ruptured in a similar manner to that of

Example 1 except that the aging temperature was set to 650°C. The rupture time of the test piece was about 1,100 hours. Then, the test piece was observed as to whether a type 4 damage has occurred or not.

5 The result was as indicated in TABLE 2.

After the rupture, precipitates and the like contained in the HAZ fine grained region of the test piece were taken on a film as they were dispersed in the material by using an extraction replica method, and
10 the carbide structure was observed under a transmission electron microscope (TEM). The extraction replica method was carried out in the following procedure.

1. The surface of the sample was polished;
2. The sample was etched with a 10%acetylacetone-
15 1%tetraammonium chloride-methyl alcohol solution by a speed method;
3. Carbon was deposited on the surface of the sample to have a thickness of 20 to 30 nm;
4. The substrate was dissolved by a similar manner
20 to the step 2 above, and thus the carbon film to which precipitates were attached was separated; and
5. The carbon film was skimmed out with a Cu mesh to make a sample to be observed.

The carbide structure of thus obtained sample was
25 observed under TEM with regard to the precipitation state of high-density agglomerated carbides, and the results obtained were shown in TABLE 2.

(Reference Example)

A simulation test was carried out in a similar manner to that of Example 1 except that the aging temperature was set to 566°C. Then, the test piece
5 (, which was not ruptured,) was observed as to whether a type 4 damage has occurred or not. In this reference example, the test was aborted after 10,000 hours.

After that, carbide structures of the piece was
10 observed as Example 3. The results were as indicated in FIG. 9 and TABLE 2. FIG. 9 is a TEM photograph showing the carbide structures under a magnification of about 500 times.

(Comparative Example 1)

15 In a similar manner to that of Example 1, a test piece (, which was not normalized,) was obtained from the HAZ fine grained region of the inner side of 90° -steel pipe elbow , which was used for about 188,000 hours as a main steam pipe of a thermal power
20 plant. The actual used conditions of this example was substantially similar to the used condition of Example 1. Then, the piece was observed as to whether a type 4 damage has occurred or not. Subsequently, as in a similar manner to that of Example 3, the carbide
25 structures of the sample were observed. The extraction replica (test piece) was taken at a depth about two thirds of the thickness from the outer surface.

The results of the observation were as shown in FIG. 3 and TABLE 2.

TABLE 2

	Aging conditions		Evaluation results	
	Temperature (°C)	Time (hours)	High-density agglomerated carbides	Type 4 damage
Example 1	680	About 220	-	Not present
Example 2	670	About 380	-	Not present
Example 3	650	About 1,100	Few	Present
Reference Example	566	10,000	Few	Not present
Comparative Example 1	-	-	Great deal	Present

As is clear from TABLE 2, in Examples 1 and 2, no type 4 damage was observed until they ruptured, i.e. they underwent an aging of even severer conditions than those actual use. The aging time of each of Examples 1
5 and 2, when converted into the actual use conditions, is about 240,000 hours. Thus, it was confirmed that the welded component of the present invention can fully withstand the long term use as a main steam pipe of a thermal power plant.

10 By contrast, a type 4 damage was observed in Comparative Example 1. It was confirmed by an observation under TEM that a great amount of high-density agglomerated carbides were created in the HAZ fine grained region of Comparative Example 1. It should
15 be noted that the agglomeration of spherical carbides proceeds, high-density agglomerated carbides are created. Typical examples of the high-density agglomerated carbides are plate-like carbides shown in FIGS. 3 to 5. The plate-like carbides are a factor
20 of the type 4 damages which occur on early stage.

As described above, in the case of the welded component of the present invention, if it is exposed to various use conditions, the high-density agglomerated carbides can be suppressed as compared to the
25 conventional welded component, thereby making it possible to reduce the type 4 damages.

In Example 3, a type 4 damage was observed;

however, the rupture of the test piece was not directly caused by the type 4 damage. It is understood that the type 4 damage can be suppressed significantly in Example 3, thereby the rupture occurred in the portion other than the HAZ fine grained region. This is because, as compared to Comparative Example 1, the amount of high-density agglomerated carbides was remarkably less in Example 3.

Meanwhile, in the case of the reference example, the base metal was normalized once as a separate step from the hot working before the welding as in the present invention. In this example, the type 4 damage was not observed even though it was aged at substantially an equal temperature and load to the actual use conditions.

However, the TEM photograph shown in FIG. 9 indicates a slight reticular carbide agglomeration has been created in a lower right section of the photograph. Further, in Example 3 as well, a few high-density agglomerated carbides were present as indicated in TABLE 2 but plate-like carbides was not observed. Here, it is considered that in the case of the above-described steel plate (refreshed STPA 24 material), another normalizing should be carried out, that is, a total of twice of normalizing should be carried out in addition to the hot working that serves also as refreshing. In this manner, a welded component

having better properties with even less high-density agglomerated carbides can be created.

By contrast, in Comparative Example 1, not only spherical carbides but also plate-like carbide
5 agglomeration regions (dark portion in FIG. 3) were present in a wide range of the sample. For the carbide agglomeration region of Comparative Example 1, the identification of precipitates was carried out by an energy dispersion X-ray analysis. In the analysis,
10 $(\text{Mo}, \text{Fe})_6\text{C}$, $(\text{Cr}, \text{Fe})_{23}\text{C}_6$ and the like were defined, which suggested that the plate-like agglomeration of carbides occurs more easily in a steel type containing a great amount of Mo and Cr, as compared to a steel type containing low contents of these.

15 As described above, according to the present invention, it is possible to provide a low-alloy steel welded component that does not easily have a plate-like agglomeration region of carbides even if it is exposed to multiple thermal cycles by submerged arc weldings,
20 and it is used under conditions where a stress created by high internal pressure is applied thereon at a high temperature. Therefore, a long-life low-alloy steel welded component that does not easily create type 4 creep damages, which are caused by a plate-like carbide
25 agglomeration region, can be obtained.

According to the present invention, it is possible to provide a long-life low-alloy steel welded component

that can be applied to a wide usage in facilities that
employs a great number of thick pipes made of ferritic
heat-resisting steel. Further, the welded component of
the present invention does not easily create creep
5 damages in a heat affected zone even if the submerged
arc welding is used.

Additional advantages and modifications will
readily occur to those skilled in the art. Therefore,
the invention in its broader aspects is not limited to
10 the specific details and representative embodiments
shown and described herein. Accordingly, various
modifications may be made without departing from the
spirit or scope of the general inventive concept as
defined by the appended claims and their equivalents.